Ontario Hydro

RESEARCH DIVISION

91602
new 82-10

ontario hydro
research division

EFFECTS OF STRAIN AGING AND RELATED PHENOMENA UPON THE TOUGHNESS OF CARBON AND CARBON-MANGANESE WELD METALS - A LITERATURE REVIEW

Report No. 86-323-K
E.T.C. Ho - Engineer
Materials Integrity Section
Metallurgical Research Department
and
O.E. Lepik - Engineer
Nondestructive and Fracture Evaluation Section
Metallurgical Research Department

ABSTRACT

A review of the technical literature has been carried out to assess the various factors affecting toughness in carbon and carbon-manganese weld metals. This survey was motivated by some test results obtained from the Darlington Leak-Before-Break Materials Test Program, which indicated that some non-stress-relieved welds in SA-106 Grade B piping material exhibited lower than expected fracture toughness at 250°C and cleavage fracture at room temperature after significant ductile tearing. The factors reviewed are strain aging embrittlement, residual stresses, dilution of weld metal and quench aging. The effect of post weld heat treatment is addressed since it is known to improve weld toughness and reduce the risk of brittle failure. Lastly, the micromechanism of cleavage fracture in steels, and the effects of microstructure and the various factors reviewed on cleavage fracture resistance, are examined.

job	file	date	report no.
83-160-552	826.232	January 14, 1987	86-323-K

91602
new 82-10

ontario hydro
research division

EXECUTIVE SUMMARY

EFFECTS OF STRAIN AGING AND RELATED PHENOMENA UPON THE TOUGHNESS OF CARBON AND CARBON-MANGANESE WELD METALS - A LITERATURE REVIEW

E.T.C. Ho - Materials Integrity Section
O.E. Lepik - Nondestructive and Fracture Evaluation Section
Metallurgical Research Department

A review of the technical literature has been carried out to assess the various factors affecting toughness in carbon and carbon-manganese weld metals. This survey was motivated by some test results obtained from the Darlington Leak-Before-Break Materials Test Program, which indicated that some non-stress-relieved welds in SA106 Grade B piping material exhibited reduced fracture toughness at 250°C and cleavage fracture at room temperature after significant ductile tearing. The factors reviewed are strain aging embrittlement, residual stresses, dilution of weld metal and quench aging. The effect of post weld heat treatment (PWHT) is addressed since such action is known to improve weld toughness and reduce the risk of brittle failure. For completeness, the micromechanism of cleavage fracture in steels, and the effects of microstructure and the various factors reviewed on a material's resistance to cleavage fracture, are examined.

Information obtained from the literature review suggests that the reduction of fracture toughness observed in some Darlington piping weld material may be attributed largely to a strain aging effect. However, since PWHT of the PN-229 weld metal did not completely eliminate the initiation of localized cleavage in some small areas of the weld microstructure, it is apparent that other factors have contributed or interacted to cause the observed cleavage fractures. The proposed material test program for characterizing the toughness of selected weld metal should help to further evaluate the necessity of PWHT in primary heat trasnport piping.

job	file	date	report no.
83-160-552	826.232	January 14, 1987	86-323-K

TABLE OF CONTENTS

		PAGE
1.0	INTRODUCTION	1
2.0	FACTORS AFFECTING WELD METAL TOUGHNESS	2
2.1	Strain Aging Embrittlement	2
2.1.1	Historical Background	2
2.1.2	The Nature of the Strain Aging Phenomenon	2
2.1.3	Dynamic Strain Aging	3
2.1.4	Interstitial Solute Concentrations Required for Strain Aging	4
2.1.5	Effect of Prolonged Aging at Service Temperatures	4
2.1.6	Implication of the Occurrence of Strain Aging and Dynamic Strain Aging on the Toughness of Carbon and Carbon-Manganese Weld Metals	5
2.2	Residual Stress Effects	6
2.3	Effects of Weld Metal Dilution	6
2.4	Quench Aging Effects	7
3.0	POST WELD HEAT TREATMENT OF WELD METAL	7
3.1	The Role of PWHT	7
3.2	The Effect of PWHT on the Toughness of Carbon and Carbon-Manganese Weld Metals	8
4.0	FACTORS AFFECTING CLEAVAGE FRACTURE RESISTANCE IN STEELS	9
4.1	The Micromechanism of Cleavage Fracture	9
4.2	The Effect of Microstructure on Cleavage Fracture Resistance	10
4.3	The Effect of Pre-Strain and Strain Aging on Cleavage Fracture Resistance	11
5.0	SUMMARY AND CONCLUSIONS	12
REFERENCES		14
TABLE 1	Results of COD Tests on 1 in (25 mm) Square Specimens of As-Deposited, Strain Aged and Thermally Stress-Relieved Weld Metals (Reference 51)	19
FIGURES		20
DISTRIBUTION		last page

LIST OF FIGURES

FIGURE 1 Effect of Residual Stress or Residual COD on the Remaining COD at (a) -25°C; (b) -77°C. (Reference 36)

FIGURE 2 The Hardness of Rimming Carbon Steel vs Aging Time at 20, 100 and 200°C (Reference 43)

FIGURE 3 Schematic Diagram showing the Various Metallurgical Changes That Can Take Place During PWHT (Reference 44)

FIGURE 4 Effect of Prestraining and Reheating at Various Stress Relieving Temperatures on Notch Toughness (A-285 and A-212 are C-Mn Steels) (Reference 46)

FIGURE 5 Recovery of Fracture Stress After Postheating of a C-Mn Steel (Reference 50)

FIGURE 6 The influence of PWHT on the Charpy V-notch Impact Strength of the Root-Run in Multi-Pass MMA Welds in QT. C-Mn-Nb-V Steel. AW-As-welded; SR-Stress Relieved (Reference 47)

FIGURE 7 Energy and Fibrosity of C-Mn Weld Metal (Electrodes of Class E6XX) Reheated for 2 h (Reference 56)

FIGURE 8 Schematic Diagram Showing Relationship Between Microstructure and Cleavage Resistance of High and Low Arc Energy Welds (Reference 8)

ontario hydro
research division

To Mr. M.R. Allen
Manager
Mechanical Design Department

EFFECTS OF STRAIN AGING AND RELATED PHENOMENA UPON THE TOUGHNESS OF CARBON AND CARBON-MANGANESE WELD METALS - A LITERATURE REVIEW

1.0 INTRODUCTION

An extensive materials test program on the mechanical properties of SA106B piping material has been carried out at Ontario Hydro Research. As described by Manning/1/, this test program is part of the Darlington leak-before-break study. Results obtained indicated that some non-stress-relieved submerged arc weld material displayed lower than expected fracture toughness at 250°C and cleavage fracture at room temperature after significant ductile tearing. However, metallurgical investigations of the weld metal (fabricated according to Ontario Hydro welding procedure PN-229) did not disclose a satisfactory explanation for these observations. The metallurgical investigations carried out did show that post weld heat treatment (PWHT) of the weld improved the toughness properties at 250 °C and the room temperature Charpy impact energy. The PWHT (2 h at 620°C) also appeared to eliminate the cleavage fracture in the fracture toughness specimens. Based on the above findings, and the tight construction schedule which did not permit an in-depth study, PWHT of welds was recommended/2/ for the existing PN-229 welds in 22" diameter and 1.37" thick piping at Darlington.

Since the ASME Code does not require PWHT of welds in material below 1.5" thickness, these results could be very significant for nuclear plant piping. There are numerous existing non-stress relieved welds in both Pickering and Bruce. These welds satisfy the impact test acceptance standards of the ASME Code. The PN-229 weld procedure will be used for the remaining piping at Darlington.

Since no satisfactory explanation has been found for the reduced toughness in the non-stress relieved welds tested, their long-term performance cannot be predicted. Recent evidence suggests that dynamic strain aging embrittlement may be responsible for the toughness reduction. Serrated flow was observed in the tensile/3/ and fracture toughness/4/ tests carried out at 250°C on the non-stress-relieved welds. This type of behaviour is observed in materials that undergo dynamic strain aging. The degree to which strain aging can affect material properties in service is not known.

A test program has been proposed/5/ that will characterize the toughness of non-stress-relieved submerged arc weld metal and determine whether the lack of PWHT in PHT piping will present problems. In support of this program, a review of the technical literature was undertaken to assess the embrittlement mechanisms which may cause reduced toughness at elevated temperatures and promote cleavage fractures of non-stress-relieved weld metals. The effect of PWHT on weld metals is also addressed since PWHT of certain weld metals is known to improve weld toughness and reduce the risk of brittle failure in service.

2.0 FACTORS AFFECTING WELD METAL TOUGHNESS

The mechanical properties of carbon and carbon-manganese weld metal, such as fracture toughness, are dependent on numerous factors. These factors can generally be divided into two basic groups: chemical compositional factors and microstructural factors. Since a detailed discussion of all the factors is beyond the scope of the present review, the reader is referred to several excellent review articles/6,7,8/ for a more comprehensive discussion on chemical and microstructural factors.

For purposes of the present review, the factors that will be examined in some details are:

1. strain aging embrittlement;
2. residual stresses;
3. dilution of weld metal;
4. quench aging.

Although the above list is by no means comprehensive, the list nevertheless covers the factors of practical importance.

2.1 Strain Aging Embrittlement

2.1.1 Historical Background

In the late 1800's it was noted that when mild steel was heated to a blue colour, it became brittle if worked/9/. This so-called "blue brittleness" is familiar to ferrous metallurgists and the phenomenon is now recognized as a manifestation of dynamic strain aging (DSA). Early investigation of the phenomenon by Le Chatelier/10/ showed that the blue brittle region was accompanied by an increase in tensile strength and that the increase in strength was a function of strain rate. Subsequently, several researchers studied in considerable detail the stress-strain curves of mild steel as a function of temperature and strain rate. The results obtained showed that within certain combinations of strain rate and temperature, serrated stress-strain curves were observed. This phenomenon of serrated yielding and flow is generally referred to as the Portevin-Le Chatelier effect, named after the original investigators of this phenomenon in aluminum alloys/11/. It is now generally accepted, as first suggested by Nabarro/12/, that serrated yielding in steel is observed when carbon or nitrogen atoms can diffuse fast enough to allow strain aging to take place during plastic deformation.

2.1.2 The Nature of the Strain Aging Phenomenon

In metals and alloys, the phenomenon of strain aging is invariably accompanied by a yield drop in the stress-strain curve. The yield drop is a manifestation of the pinning and the subsequent release of dislocations already present in the metal or alloy by certain solute atoms. This segregation of suitable solute atoms to the dislocations occurs because the process reduces the overall elastic strain energy of the system. In general, the amount of solute needed to pin the dislocations is very small. For example, in an annealed metal, only about one atomic site in 10^8 is at the core of a dislocation. This density rises to about one site in 10^3 in a heavily worked metal. Therefore, it is evident that badly misfitting solute atoms that are mobile at relatively low temperatures can saturate dislocations even when present in very low concentrations.

The above considerations underscore the dramatic effect of small amounts of carbon or nitrogen on the mechanical properties of carbon and low-alloy steels. Since interstitial carbon and nitrogen atoms are very mobile in body centered cubic (bcc) iron, even at ambient temperatures, and since they strongly distort the iron lattice, conditions are extremely favorable for intense segregation of these atoms to all dislocations in bcc iron or ferrite. This immobilization of dislocations by the strong pinning effect of the interstitial solute atoms means that the density of the dislocations available for glide (ie, further straining) is reduced virtually to zero until very high applied stresses are reached. At high applied stresses, the dislocations can either break away from their segregated atoms or "Cottrell atmospheres" or be created at sites of stress concentration.

The above-described solute-dislocation interactions in steels result in flow curves containing upper and lower yield points. This type of yielding also leads to a plastic instability known as the "yield elongation" or "Luders strain", in which the metal has to strain to a certain amount, usually up to about 0.05, before work hardening can increase its flow stress sufficiently to overcome the initial loss in strength due to yielding, to produce a rising stress-strain curve. The instability propagates along the specimen as a Luders band or stretcher strain. In a homogeneous metal, the Luders front moves at a constant speed, at constant stress/13/. The lower yield stress thus remains approximately constant during the yielding elongation, although it may jump from one level to another as Luders bands form in the unyielded regions.

If the yielded sample is unloaded and immediately reloaded, it exhibits no yield point since it contains unpinned dislocations. However, if the sample is rested or mildly heated to permit the interstitial atoms to migrate to the dislocations again, the yield point returns. For mild steel, strain aging usually takes a few days at room temperature, or about 30 min at 100°C. The reappearance of the yield point is controlled by the diffusion of carbon and nitrogen atoms which require an activation energy of 0.8-0.9 eV. However, if the aging treatment is more prolonged, there may also be some general hardening caused by precipitation of carbides and nitrides.

In essence, the embrittlement due to strain aging may be explained by the pinning of the disolocations by the interstitial solute atoms. This locking-up of the dislocations results in high work-hardening rates which allow very high stresses to be achieved at low strains. The high work-hardening also leads to low ductility because the onset of plastic flow instability or necking occurs at small strains.

2.1.3 Dynamic Strain Aging

The micromechanism of DSA is similar to that of strain aging except that the pinning of the dislocation occurs during plastic deformation. Although there is still considerable debate as to what the precise details of the micromechanism of DSA are/14-16/, there is nevertheless general agreement that the phenomenon involves the interaction between moving dislocations and diffusing solute atoms. In general, the occurrence of DSA in iron and steel is manifested by serrated stress-strain curves, high work-hardening rates, negative temperature and strain-rate dependence of the flow stress and a reduction in ductility/17/. It has been found that the activation energy for the onset of serrations is comparable to that of interstitial diffusion in α-iron. Furthermore, the activation energy for the disappearance of the serrations can be interpreted as the sum of the interstitial diffusion energy and the binding or interaction energy between dislocations and interstitial solutes.

The occurrence of DSA in carbon and low alloy steels results in a degradation of mechanical properties and may lead to embrittlement. Macroscopic manifestations of the embrittlement are:

1. low ductility in tensile tests;
2. low fracture toughness properties in crack growth resistance (J_R) tests;
3. increased transition temperatures in Charpy impact tests.

Serious degradation of materials properties may lead to fast or cleavage fracture under certain conditions.

2.1.4 Interstitial Solute Concentrations Required for Strain Aging

It has been suggested/18/ that strain aging in mild steel may largely be attributed to nitrogen with a smaller contribution from carbon. The main reason for this hypothesis is that nitrogen is much more soluble than carbon in ferrite at low temperatures. For temperatures less than 100°C, the solubility of nitrogen is about 100 times greater than that of carbon. As noted earlier, relatively small amounts of interstitial carbon and/or nitrogen can cause strain aging. For example, a nitrogen content of 0.0001% can produce detectable aging, while 0.001 to 0.002% can produce severe aging/19/.

Nitrogen levels of this magnitude can be present in weld metals. In manual metal arc (MMA) welds, cellulosic and basic electrodes have total nitrogen levels of around 0.004 - 0.014% by virtue of the heavy shielding effect afforded from the decomposition of the cellulose and basic carbonates respectively. By comparison, acid and rutile coated electrode MMA welds have typically higher mean nitrogen levels (0.022-0.050%) which is attributable to the longer arc lengths and higher oxygen levels/20/. Nitrogen levels may also be locally increased in the root region of MMA welds owing to the difficulty in shielding the root pass. With the submerged arc process, weld metal nitrogen levels vary mainly between 0.004-0.014%/21/. It must be noted that the above-mentioned nitrogen levels are the total amount in the weld metal, which is comprised of the dissolved and combined nitrogen. The dissolved or free nitrogen content responsible for strain aging does, however, tend to increase with a higher total nitrogen content.

The deposited weld metal in the as-welded condition can exhibit a large sensitivity to strain aging because a high free nitrogen/carbon content can be maintained in solid solution by the normal cooling rates in a welding operation. The higher the supersaturation the more pronounced is the process of strain aging. The usual means of reducing strain aging is by lowering the concentration of free nitrogen and carbon in the ferrite. Free nitrogen is reduced by incorporating nitride forming elements such as aluminum, titanium and silicon while interstitial carbon levels can be reduced by slow cooling and by elements that form stable carbonitrides and carbides.

2.1.5 Effect of Prolonged Aging at Service Temperatures

The natural aging times of welds are of the order of years at service temperatures. For this reason fracture toughness tests carried out within several weeks of welding, without artificial aging, may result in overly optimistic results.

To assess the changes in fracture toughness during exposure to service conditions several investigators have studied the effect of prolonged aging under stress on a quenched and tempered PWR pressure vessel steel (A533B). The results of these tests would generally be applicable to other steels such as C and C-Mn steels. The results from one study/22/ indicated that an aging treatment under stress for a period of up to four years after prior prestraining did not significantly reduce the fracture toughness over and above that resulting from dynamic strain

aging. Another study/23/ showed that the toughness of A533B plate material and submerged arc weldment in the PWHT condition was not significantly affected at 290°C for aging periods up to 10^4 h. These results tend to confirm that much of the reduction in toughness occurs after very short aging times at elevated temperatures, and that prolonged aging in the service temperature range will not result in any further significant reduction in toughness.

2.1.6 Implication of the Occurrence of Strain Aging and Dynamic Strain Aging on the Toughness of Carbon and Carbon-Manganese Weld Metals

Several studies/24-26/ have shown that local strains caused by thermal and mechanical cycling during the course of a welding operation can lead to embrittlement and deterioration of materials properties in the close vicinity of the joint. On the microscale, this deterioration in materials properties frequently involves the pinning of dislocations by mobile solute atoms, which are generally carbon and nitrogen atoms. The manifestations of this pinning are the occurrences of strain aging and DSA phenomena in the welded material.

Brindley/27/ has found that deformation under DSA conditions facilitates easier void link-up and increases the rate of carbide cracking in mild steel. He also reported that the main effect of DSA on the ductile fracture process is the reduction in ductility which directly results from the onset of necking at small strains brought about by a change in the work hardening characteristics. Berkhout and van den Brink/28/ observed that strain aging has a dual effect on the transition temperature curve of some welds: a decrease in the upper shelf energy combined with a shift of the curve to higher temperatures. Gerberich et al/29/ proposed that a subcritical flaw may grow in service by a slow crack growth mechanism attributable to "spontaneous" strain aging. This mechanism is very similar to the delayed hydride cracking (DHC) mechanism reported/30/ for Zircaloy-2 and Zr-2.5 Nb pressure tubes in CANDU power reactors. According to Gerberich et al, the preferential precipitation of carbides in the plastic zone of the crack tip embrittles this zone. After a sufficient time for precipitation, the crack propagates through the embrittled zone and arrests in material that has a much lower density of carbides. This cycle repeats as the material slowly cracks.

Chakravartty et al/31/ studied the effect of DSA on the materials properties of A203D nuclear structural steels. They found that the microstructure of the steel had minimal influence on the changes in the mechanical properties during deformation under DSA conditions. These workers showed that for the steel investigated, the adverse effects of strain aging or DSA cannot be reduced appreciably by control of the materials' microstructure. They concluded that reduction of the detrimental effects of strain aging or DSA on the mechanical properties requires chemistry control of the material. In a review of the factors which affect toughness in welded steels, Kerr/6/ reported that the occurrence of strain aging due to interstitial nitrogen or oxygen in the weld zone has a large effect on the toughness of the steel.

In a study of the aging of steels and its relation to weldability, Vrtel/32/ stated that weldability is synonymous with the capacity of the metal to accommodate without cracking very high local strains in local regions of complex phase transformations and stress concentrations. He reported that considerable deterioration of properties of weldable steels can result from the combined effects of deformation and aging. The author stated that practical experiences gained from analyses of service failures indicate that the critical stress for fracture is affected significantly by

the geometry of the weld and by the local embrittlement of the heat-affected zone (HAZ). In particular, a brittle fracture can only occur when the strain in the embrittled zone reaches a critical value which depends on the material and the temperature.

The effects of strain aging and DSA embrittlement in pressure vessel steels have been studied by various researchers/7,33-35/. Formby and Charnock/33/ found that a brittle fracture can be induced by service pressurization in the DSA temperature regime for the C-Mn steel studied, if a suitable crack was present. They also reported that a catastrophic vessel failure can only occur if the conditions necessary to maintain a fast-running crack in the non-embrittled material were fulfilled. Sandberg/22/ studied the influence of strain aging on the fracture toughness in a A533B pressure vessel steel since earlier work/34/ had demonstrated that the steel was susceptible to DSA. The fracture toughness of the steel was found to decrease with increasing temperature over the range 100-300°C. This decrease in fracture toughness can be correlated with the occurrence and the magnitude of the DSA effect observed in the material.

2.2 Residual Stress Effects

Residual stresses arise in welded components from the localized shrinkage associated with the cooling of hot weld metal that is restrained by the cold sections of parent metal. Tensile residual stresses of yield point magnitude are generated in the welding direction/35/. The residual stress distribution in a weld is complex and can vary depending on the nature of the joint, the degree of external restraint, the number of passes involved, and to some extent the welding parameters used. The role of residual stresses in any given component can only be accurately assessed in a full scale test, since such stresses will be largely removed if the weldment is cut up into smaller pieces such as those used for impact or fracture toughness tests.

Tensile residual stresses can cause a significant reduction in resistance to brittle fracture. Akita et al/36/ conducted tests to investigate the effects of residual stress on brittle fracture initiation of welded and preloaded plates. Their analysis considered the total crack opening displacement (COD) to be composed of a COD due to applied load and a COD due to residual stress. The results presented in Figure 1 indicate that for plates with higher residual COD values from higher welding or preloading residual stresses the remaining COD to cause fracture decreases until a constant value is attained. Other studies have shown residual stresses to vary from having a slight or no effect on the fracture loads at high toughness to being the sole cause of fracture/37,38/. If lower shelf toughness is approached and residual stresses are known to be present, then they should be considered with the loading stresses if linear elastic behaviour is anticipated. On the other hand, if the steel exhibits upper shelf toughness or sufficient ductility to allow substantial plasticity during fracture, then the influence of residual stress on fracture becomes negligible.

2.3 Effects of Weld Metal Dilution

The toughness of multipass welds in C and C-Mn steel plates can be affected by low toughness in the root region. The phenomenon of poor fracture toughness in the root region is well known and is partly attributed to high levels of weld metal dilution from parent plate material/8/. Parent plate dilution by carbon and microalloying elements such as Nb, V and Al, present as either microalloying elements or residual impurities, can lead to embrittlement as a result of changes in the microstructure.

Carbon causes the yield and tensile strengths to increase and ductility to decrease in as-deposited weld metals. These changes with increasing carbon level are associated with an increasing carbide volume fraction which decreases cleavage resistance/39/. Precipitation of grain refining elements, such as Nb and Al from the parent plate, in the root area has been reported to have a deleterious effect on the microstructure and toughness/40,41/. Dilution effects from microalloying elements, however, are not expected to affect the performance of the root region at the levels normally present in C and C-Mn steels/42/.

2.4 Quench Aging Effects

Quench aging arises when low-carbon steels are rapidly cooled from sub-critical temperatures. Rapid cooling down to room temperature from temperatures between 580°C (at which the solubility of nitrogen in α-iron is highest) and 723°C (where the solubility of carbon in α-iron is highest) results in the formation of a supersaturated metastable interstitial solid solution of nitrogen and carbon in ferrite. Upon holding at room temperature iron nitride and carbide precipitate from the supersaturated ferrite and precipitation or quench-age hardening occurs. Figures 2 illustrates how the hardness of a rimming carbon steel changes with time when quench-aged at 20, 100 and 200°C/43/. This hardening can cause the transition temperature to increase significantly.

The phenomenon of quench aging is largely avoided in welds because the cooling rates from welding are much slower than those from quenching and the effect can be eliminated by overaging at temperatures in excess of 100°C.

3.0 POST WELD HEAT TREATMENT OF WELD METAL

3.1 The Role of PWHT

The primary purpose of PWHT at temperatures below the critical or recrystallization temperature is the relief of residual stresses. Stress relaxation by creep processes occurs during PWHT which converts elastic strains to plastic strains thereby reducing stresses to a level just below the yield point of the material at the temperature of the heat treatment. For this treatment carbon steels are held at temperatures typically ranging between 595 and 675°C for a minimum of 1 h per inch of section thickness. PWHT also improves toughness and mechanical properties, aids the removal of hydrogen, increases dimensional stability and reduces the potential of stress corrosion cracking in certain environments.

Various metallurgical changes can occur during the course of PWHT that can influence the mechanical properties of welded structures. The favourable effects may be summarized as follows:

(i) tempering of the hardened structures within the HAZ and weld metal,
(ii) restoration of the properties embrittled by strain, and
(iii) restoration of the properties in zones embrittled by heating during the welding process.

PWHT can also cause the following unfavourable effects in some materials:

(i) reduction in the tensile strength of the parent metal,
(ii) embrittlement of the parent metal, HAZ or weld metal,
(iii) reheat cracking in the HAZ.

3.2 The Effect of PWHT on the Toughness of Carbon and Carbon-Manganese Weld Metals

The various factors tending to increase or decrease weld metal toughness during PWHT have been summarized in model form by Dolby/44/. This model is similar to the weld toughness model first developed by Garland and Kirkwood/45/. The net effect of the various factors, shown in Figure 3, determines how the toughness is affected following PWHT. PWHT tempers martensitic regions into a ferrite/carbide aggregate, which has a markedly less deleterious effect on weld metal cleavage resistance, and removes carbon from solid solution. Factors that are detrimental to toughness include precipitation hardening by microalloying elements (if present), and the decomposition of tough retained austenite regions to give regions containing coarse carbides.

A major effect of PWHT in C and C-Mn weld metals is the restoration of properties after strain embrittlement. Numerous investigations have shown that a stress relief heat treatment at temperatures above approximately 500°C alleviates strain aging embrittlement/46-51/. A number of these studies have determined the strain aging characteristics of parent metal in various heat treated conditions by testing artificially strain aged material. The specimens are typically strained by 5%, and are then heated in the temperature range at which aging is most severe (~250°C). In one such investigation the Charpy notch toughness of various strain aged C-Mn steels showed marked improvement at a stress relieving temperature of 620°C (Figure 4)/46/. Tests of this type can be used to infer the behaviour of strain-aged weld metal which, as noted earlier, is particularly susceptible to strain aging embrittlement in the root and central regions of a multipass weld. Although it is difficult to establish whether strains of the order of 5% can actually occur during welding, work on the dislocation sub-structures produced after dynamic strain aging/52/ has indicated that a 5% strain at 250°C is necessary to produce the level of dislocation density observed in the root region of multi-pass welds/47/.

Wells/48/ was the first to reproduce systematically low load brittle fractures in the parent metal of welded wide plates of conventional C-Mn steel and confirm the beneficial effect of a furnace stress relief. In these tests a notch is placed in the edge of the parent plate that is to be welded. Strain aging embrittlement occurs at the notch tip as it is subjected to the thermal and plastic strain cycles from welding. Results showed that low stress fractures were prevented at the as-welded transition temperature for the specimens stress relieved at 450, 500 and 650°C and that the ductility increased with an increase in stress relieving temperature. Soete/49/ also used the Wells wide plate test to show that strain-aged specimens of C-Mn steel recovered when heated at temperatures greater than 500°C.

In a separate study, Toyooka and Terai/50/ showed that the low fracture stress values of as-welded, longitudinally welded and center notch wide-plate tensile specimens of C-Mn steel were recovered by PWHT at short soaking times, as illustrated in Figure 5. Statistical methods were also used to show that the recovery of fracture toughness appeared to have no distinct relationship with the residual stress values and that the recovery of high fracture stress values were highly correlated with the reduction of hardness of the strain-hardened specimens.

In studies on weld metals Cochrane et al/47/ showed that the toughness of the root region in multi-pass MMA welds in high strength steel plate is controlled by its DSA response. The application of a PWHT in excess of 600°C was beneficial to root properties as detailed in Figure 6. Dawes/51/ carried out COD tests on strain-aged MMA weld metals and reported that a PWHT of 600°C for 1 h removed the strain aging damage. The PWHT, however, generally introduced concomitant embrittling effects which prevented full recovery of the as-deposited properties (Table 1).

The results of these investigations show that a thermal stress relief can remove to a large extent strain aging damage and induce a large recovery of the initial toughness in C and C-Mn steels and weld metals. The chief reason for this effect is that the heat treatment reduces the concentration of carbon and nitrogen in solution in ferrite. PWHT at temperatures of 600°C is sufficient to precipitate nitrides and carbonitrides by combining with Al, Ti, V and Si, thus lowering the free interstitial content/53,54/. Another reason for the improvement in toughness is attributed to recovery, since the dislocation density is known to decrease with increasing temperature/55/.

While a PWHT of C and C-Mn weld metals has generally been observed to be beneficial a number of studies have demonstrated a deleterious effect from PWHT. A factor that can act to reduce toughness in MMA deposits, which commonly contain some Ti, is precipitation hardening by TiN/8/. Welds made at low heat input can also reduce toughness after PWHT. Wheatley and Baker/56/ reported that low heat input welds contained greater proportions of retained austenite which decomposed after PWHT to give brittle carbides. A significant deterioration in impact properties was observed, with the transition temperature rising by about 50°C, after heating at 650°C (Figure 7).

The formation of grain boundary carbides can lead to stress relief embrittlement. Farrar et al/57/ found that the critical COD transition temperature increased by 30-60°C when plain carbon steel weld metal containing 1.0% Mn and 0.08% C was stress relieved at 625°C for 4 h. In this investigation the stress relief promoted the formation of grain boundary carbide films which offer easy paths for crack propagation. Similar results were reported by Tuliani/58/, who observed that with C-Mn weld metals containing <0.8% Mn the effect of PWHT was to raise the Charpy transition temperature, whereas with weld metals containing >0.8% Mn the transition temperature was lowered by PWHT. It is thought that in weld metals with a lower manganese content a localized carbon-concentration profile develops during the austenite → pro-eutectoid ferrite transformation which causes grain boundary carbide films to form after stress relief.

4.0 FACTORS AFFECTING CLEAVAGE FRACTURE RESISTANCE IN STEELS

4.1 The Micromechanism of Cleavage Fracture

The most widely accepted criterion for cleavage fracture in mild steel is that failure occurs when some critical tensile stress, σ_f, is locally exceeded within the material. Cleavage fracture results in the creation of new free surfaces by separation across well defined crystallographic planes in the bcc lattice. In ferritic steels, these are the cube or {100} planes. Analyses of the stress field ahead of a crack tip at the atomistic level suggest that in bcc iron, yielding will probably occur at a loaded crack tip before the crack can extend by cleavage. The above result implies that some form of stress elevation mechanism must operate for cleavage to occur in steels. Studies have shown that plastic flow is a necessary precursor to cleavage and that yielding, which may be by slip or twinning, is involved in the nucleation of cleavage fracture. It has also been shown that if

the fracture is slip-initiated, then the σ_f value is essentially independent of test temperature and applied strain rate. However, if the fracture is initiated by a twinning process, then σ_f is very sensitive to temperature and strain rate/59/. The above temperature and strain rate dependence arises because the shear stress needed to form the twinning dislocations is dependent on temperature and strain rate.

As described by Curry/60/, several micromechanisms of cleavage fracture have been proposed in the literature. Although the various models differ somewhat in detail, they all involve the formulation of an energy balance criterion for cleavage which is very similar to the first one of this kind proposed by Griffith/61/ to explain why the fracture strength of real solids is significantly lower than the ideal strength. Griffith's energy criterion is as follows:

$$\sigma^2 = \frac{\pi}{2(1-\nu^2)} \quad \frac{E\gamma}{c}$$

where ν is Poisson's ration, E is Young's modulus, γ is the surface energy of the solid and c is the radius of a very thin circular crack in a plane normal to the tensile stress, σ. Griffith's criterion states that the crack will not extend unless the work done, by forces applied to the body, is at least equal to the increase of surface energy of the solid. Most of the proposed cleavage micromechanisms also invoke a dislocation pile-up mechanism for stress elevation. One of the more widely used criteria is that due to Smith/62/. This criterion was developed for the case where the crack nucleus is initiated in a grain boundary carbide of thickness C_o by a dislocation pile-up of length equal to half the grain diameter, d. The criterion is as follows:

$$\left(\frac{Co}{d}\right) \sigma_f^2 + \tau_{eff}^2 \left\{1 + \frac{4}{\pi}\left(\frac{Co}{d}\right)^{1/2} \frac{\tau_i}{\tau_{eff}}\right\}^2 \geq \frac{4E\gamma_p}{\pi(1-\nu^2)d}$$

where τ_i is the lattice friction stress, τ_{eff} is the effective shear stress, γ_p is the effective surface energy of ferrite, and E and ν have their usual meanings. The above criterion emphasizes the role of yielding parameters in determining the fracture stress, as well as the influence of carbide thickness. Work carried out by Curry and Knott/63,64/ shows that for cleavage fracture to occur, the attainment of a critical tensile stress is not a sufficient condition. In particular, since crack tip blunting prevents a sharp crack from propagating when the energy balance is first satisfied, the cleavage fracture stress has to be exceeded over some microstructurally determined characteristic distance ahead of the crack tip before cleavage fracture could occur.

4.2 The Effect of Microstructure on Cleavage Fracture Resistance

From the preceeding discussion, it is apparent that the microstructure of a material plays an important role in determining its cleavage fracture toughness. Results from several studies indicate that σ_f is strongly dependent on grain size, with σ_f being higher in fine grained materials. This explains in part the different cleavage resistances of welds made by various welding processes. Submerged arc welds generally tend to have a coarser microstructure than MMA welds, and consequently a lower resistance to cleavage, as illustrated in Figure 8/8/, because higher heat inputs are used in their manufacture.

Cleavage fracture can occur in steels which contain no discrete carbide particles, such as lath martensite and low-C bainites. In these types of microstructures, cleavage is controlled primarily by the martensite or bainite packet size, although the lath width has also been reported to have an influence. Many studies/63-65/ have shown that the morphology and the size and spatial distribution of carbide particles play an important role in determining the size of the "characteristic distance" which control cleavage fracture in steels. In particular, these studies have shown that the characteristic distance is determined by statistical competition between crack nuclei (ie, carbide particles) of different sizes. Thus, the more numerous and finer the carbides, the higher the local fracture stress and hence the tougher is the material to cleavage fracture. The above analyses of the role of carbide particles also applies to other non-metallic inclusions which can act as cleavage crack nuclei. Therefore, any microstructure which contains locally weak regions such as grain boundary ferrite, containing non-metallic inclusions, will facilitate or promote cleavage fracture.

4.3 The Effect of Pre-Strain and Strain Aging On Cleavage Fracture Resistance

The effect of pre-strain and strain aging on cleavage fracture resistance has been investigated by many researchers/66,67/. Lindley/66/ reported that a strong strain aging treatment (1 h at 200°C) induces brittle failure in a high manganese steel in testing at -196°C. However, if the fully strain-aged material were given a re-strain of 3% at 20°C prior to low temperature testing, a ductile failure mode was re-established. The suppression of twinning by the 20°C re-strain appeared to have played a key role in preventing brittle fracture. Groom and Knott/67/ reported that σ_f was increased and that twinning was suppressed by a prestrain in a mild steel. They attributed the increase in σ_f to the need to achieve a critical displacement in a material of increased flow stress. However, these workers also noted that the prestrain increased the temperature at which the fracture coincides with general yielding.

Since the thermomechanical cycles induced by the deposition of filler passes are known to cause strain aging in weld metal, especially the root region, the influence of strain aging on the fracture characteristic of weld root material has been investigated by several workers/47,68/. McRobie and Knott/68/ found that static strain aging reduces markedly the toughness of both as-deposited and reheated weld microstructures. After the most severe strain aging treatment investigated (three successive treatments of 2% nominal plastic strain at room temperature following by aging for thirty minutes at 230°C), a transition from fibrous to cleavage crack growth was observed in COD testing at -10°C. In a comprehensive study of root-run toughness in multipass MMA welds, Cochrane et al/47/ reported that the toughness of the root region was mainly controlled by its DSA response to the thermomechanical cycle imposed by the welding process. Their study indicates that procedural changes, such as welding heat input control, offer the greatest potential for the improvement of root-run toughness.

In summary, the degradation in toughness properties with increasing severity of strain aging can be correlated with the known effects of strain aging on the material's flow properties, namely:

a) large increases in yield strength due to pinning of mobile dislocations;
b) an increase in the temperature at which the fracture stress is coincident with the general yield stress;
c) an upward shift in the brittle-ductile transition temperature.

5.0 SUMMARY AND CONCLUSIONS

A review of the technical literature on the factors affecting toughness in C and C-Mn weld metal has been carried out. The factors reviewed include strain aging embrittlement, residual stresses, dilution of weld metal and quench aging. Available information indicates that conditions present and processes occurring during welding exacerbate strain aging embrittlement. To minimize such embrittlement, the free interstitial content, mainly carbon and nitrogen, in weld metal must be kept at very low levels. This can be achieved by minimizing the absorption of nitrogen during the welding operation and by tying up the interstitials in the form of carbide or nitrides. The residual stress distribution in a weld is a very complex function of the joint geometry, the degree of external restraint, and to some extent the welding parameters. Residual stresses generally reduce toughness by permitting strain aging to occur and also by decreasing the level of applied stresses needed to reach critical stresses for cleavage initiation. In most cases, its detrimental effect on toughness can be eliminated or minimized by an appropriate PWHT. Poor fracture toughness in the root region of a multi-pass weld has been linked to high levels of weld metal dilution from parent plate materials. The loss of fracture toughness is generally attributed to a local increase in carbon content and to precipitation of brittle second-phase particles in the root region. However, dilution effects from microalloying elements are not expected to have a detrimental effect on the toughness of the root region at the levels normally present in C and C-Mn steels. Quench aging is caused by rapid cooling of the weld which leads to the formation of supersaturated interstitial solid solutions of C and N in ferrite. Upon holding at low temperatures, precipitation of coherent second-phase particles of nitride and carbide results in significant hardening and reduction in fracture toughness. Quench aging is generally avoided in welds because the cooling rates associated with normal welding processes are too low for supersaturation to occur.

PWHT of welds is generally beneficial because it reduces residual stresses, tempers hard microstructures such as martensite and upper bainite, and allows recovery processes to occur which restore the properties of material embrittled by strain. However, under certain situations, PWHT may cause unfavourable effects in the weld metal such as reduction in tensile strength and embrittlement.

Many features of the fracture behaviour of ferritic steels can be explained by applying the criterion that the critical tensile or cleavage fracture stress must be exceeded over some microstructurally determined characteristic distance before cleavage fracture can occur. The dependence of cleavage fracture resistance on microstructure is embedded in the characteristic distance parameter. This parameter is determined by statistical competition between crack nuclei of different sizes. Strain aging lowers a material's resistance to cleavage fracture because of its known detrimental effects on the material's flow properties, such as large increases in yield strength.

A comparison of the various embrittlement mechanisms with the results of the Darlington Leak-Before-Break Materials Test Program strongly suggests that the primary mechanism of embrittlement is strain aging/69/. Examples from the literature have clearly shown that strain aging decreases toughness at elevated temperatures, particularly in the temperature range where dynamic strain aging is most active, and lowers markedly the cleavage resistance of as-deposited weld metal. The results of the test program show similar behaviour. Fracture toughness tests of weld metal specimens, welded by use of Ontario Hydro weld procedures not requiring PWHT (PN-107 and PN-229) exhibited serrated load displacement curves at 250°C, while two stress relieved PN-229 welds showed no serrations. Where signs of DSA are present, the fracture

toughness represented by the crack growth resistance or J_{Ic}, is lower. In addition, the as-received PN-229 welds showed cleavage fracture at room temperature after significant ductile tearing. To further quantify the above observations, a test program has been proposed/5/ which will characterize the effects of strain aging on cleavage resistance, and fracture toughness at elevated temperatures, in the weld metals to be investigated.

It is important to note, however, that all the effects observed in the Darlington test program cannot be attributed solely to strain aging. A PWHT, which primarily has beneficial effects on C and C-Mn weld metals, did not prevent the initiation of localized cleavage. A recent examination of the fracture surface of a PN-229 weld metal specimen that had received a PWHT clearly showed small sites of cleavage fracture within the region of ductile tearing. The cleavage sites had initiated in weld passes which appeared to have a coarser microstructure as seen from the post-test fracture surface. This behaviour confirms information supported in the literature that resistance to initiation of cleavage fracture is principally dependent on the scale of the microstructure. Residual stresses may also have influenced the fracture behaviour of the weld metal. The presence of high tensile residual stresses are known to increase the risk of brittle fracture. However, the magnitude of the residual stresses, by the process of relaxation, may have been lowered in the as-welded small specimens to the point where their role in promoting cleavage has been substantially reduced. It is apparent, therefore, that a number of factors have contributed or interacted to cause the observed cleavage fractures.

Recommended by:

R.G. Fleck
Supervising Engineer
Materials Integrity Section
Metallurgical Research Department

Prepared by:

E.T.C. Ho
Engineer - Metallurgical
Materials Integrity Section

J.A. Baron
Supervising Engineer
Nondestructive & Fracture
Evaluation Section
Metallurgical Research Department

O.E. Lepik
Engineer
Nondestructive & Fracture
Evaluation Section

Approved by:

for J. Brown
Manager
Metallurgical Research Department

ETCH/OEL:ljh

REFERENCES

1. B.W. Manning, "Fracture Mechanics Leak-Before-Break Material Test Program", Design and Development - Generation Division Report No. 85120, September 1985 (Rev. 2).

2. M.L. Lau, Memorandum to M.A. Machacek, "Darlington GS - Post Weld Heat Treatment of N.P.S. 22", SG Inlet Line", File No. 903-NK38-03556P, April 14, 1986.

3. E.T.C. Ho and M.J. McGraw, "Tensile Properties of ASME SA106 Grade B and SA105 Darlington Primary Heat Transport Piping Material at 293 and 523 K", Ontario Hydro Research Division Report No. 86-11-K, July 1986.

4. B. Mukherjee and D.W. Carpenter, "Darlington Leak Before Break Material Test Program", Ontario Hydro Research Division Report, to be issued.

5. J. Brown, Memorandum to J.T. Martin, "Research Proposal: Effect of Post Weld Heat Treatment on Toughness of Submerged Arc Weldments in SA106B Piping Material", OHRD File 823.121 (T), July 15, 1986.

6. H.W. Kerr, "A Review of Factors Affecting Toughness in Welded Steels", Int. J. Pres. Ves & Piping, Vol 4, No. 2, April 1976, pp. 119-141.

7. K.E. Dorschu, "Factors Affecting Weld Metal Properties in Carbon & Low Alloy Pressure Vessel Steels", Welding Research Council Bulletin No 231, October 1977.

8. R.J. Pargeter and D.J. Abson, "Factors Influencing the Toughness of Weld Metals in Pipeline, Structural, and CMn Pressure Vessel Steels", Int. Conference on 'Fitness for Purpose Validation of Welded constructions', London, November 1981, Paper 42.

9. C.E. Stromeyer, Minutes of Proc. of Inst. of Civil Engr., Vol. 80, 1885, p 114.

10. F. Le Chatelier, Revue de Metallurgie, Vol 6, 1909, p 915.

11. A. Portevin and F. Le Chatelier, Compt. Rend. Acad. Sc., Vol 176, 1923, p 507.

12. F.R.N. Nabarro, "Report on Strength of Solids", London, Physical Soc., 1948, p 38.

13. A.H. Cottrell, "An Introduction to Metallurgy", Edward Arnold Ltd., London, 1967, p 391.

14. A. Van den Beukel, "On the Mechanism of Serrated Yielding and Dynamic Strain Aging", Acta Metallurgica, Vol 28, No. 7, July 1980, pp 965-969.

15. A. Van den Beukel and U.F. Kocks, "The Strain Dependence of Static and Dynamic Strain-Aging", Acta Metallurgica, Vol 30, No. 5, May 1982, pp 1027-1034.

16. U.F. Kocks, R.E. Cook and R.A. Mulford, "Strain Aging and Strain Hardening in Ni-C Alloys", Acta Metallurgica, Vol 33, No. 4, 1985, pp 623-638.

17. A.S. Keh, Y. Nakada and W.C. Leslie, "Dynamic Strain Aging in Iron and Steel", in Dislocation Dynamics, A.R. Rosenfield, G.T. Hahn, A.L. Benent, Jr., R.I. Jaffee (Eds.), McGraw-Hill, 1968, pp 381-408.

18. J.D. Baird, "The Effects of Strain-Aging Due to Interstitial Solutes on the Mechanical Properties of Metals", Metallurgical Reviews, Vol 16, February 1971, pp 1-18.

19. M.M. Salama, "An Assessment of Strain-Aging Effect on Weldable Offshore Structural Steel", J. of Petroleum Technology, Vol 36, No. 1, January 1984, pp 141-146.

20. P.D. Blake, "Nitrogen in Steel Weld Metals", Metal Construction, Vol 11, No. 4, April 1979, pp 196-197.

21. P.D. Blake, "Oxygen and Nitrogen in Weld Metal", Welding Research International, Vol 9, No. 1, 1979, pp 23-57.

22. O. Sandberg, "Influence of Strain Aging on Fracture Toughness of the Pressure Vessel Steel A533B at Elevated Temperatures", SMIRT-8, Brussels, Belgium, August 19-23, 1985, Paper F2 1/2, pp 409-416.

23. S.G. Druce, G. Page, G.R. Jordan and J.A. Hudson, "Effect of Thermal Aging on Mechanical Properties of PWR PV Steels and Weldments", SMIRT-8, Brussels, Belgium, August 19-23, 1985, Paper F2 1/1, pp 401-408.

24. C.F. Tipper, "Material Damage Attributable to Welding: A Commentary", British Welding J., Vol 13, No. 8, August 1966, pp 461-466.

25. C. Mylonas, D.C. Drucker and J.D. Brunton, "Static Brittle-Fracture Initiation at Net Stress 40% of Yield", Welding J., Vol 37, No. 10, October 1958, pp 473 s-479 s.

26. G.M. Boyd, "Metallurgical Concomitants of Residual Stress", British Welding J., Vol 8, No 7, July 1961, pp 343-348.

27. B.J. Brindley, "The Effect of Dynamic Strain-Aging on the Ductile Fracture Process in Mild Steel", Acta Metallurgica, Vol 18, No 3, March 1970, pp 325-329.

28. C.F. Berkhout and S.H. van den Brink, "How Strain Aging Influences the Toughness of Some Weld Metals", Welding and Metal Fabrication, Vol 46, No. 5, June 1978, pp 347-351.

29. W.W. Gerberich, C.E. Hartbower and P.P. Crimmins, "Spontaneous Strain Aging as a Mechanism of Slow Crack Growth", Welding J., Vol 47, No. 10, October 1968, pp 433 s - 443 s.

30. E.C.W. Perryman, "Pickering Pressure Tube Cracking Experience", Nuclear Energy, Vol 17, No 2, April 1978, pp 95-105.

31. J.K. Chakravartty, S.L. Wadekar, T.K. Sinha and M.K. Asundi, "Dynamic Strain Aging of A203D Nuclear Structural Steel", J. Nuclear Materials, Vol 119, No 1, November 1983, pp 51-58.

32. J. Vrtel, "Aging of Steels and its Relation to Weldability", British Welding J., Vol 15, No. 10, October 1968, pp 501-508.

33. C.L. Formby and W. Charnock, "The Effect of Dynamic Strain-Aging Embrittlement on Pressure Vessel Integrity", Institution of Mech. Engrs. Conf. on "Practical Application of Fracture Mechanics to Pressure Vessel Technology", April 1971, Paper C1/71, pp 1-7.

34. B. Ostensson and R. Westin, "The Fracture Toughness of A533B Pressure Vessel Steel at Low Strain Rate", S-573 AB Atomenergi, 1977.

35. A.J.A. Parlane, "Origin and Nature of Residual Stresses in Welded Joints", in 'Residual Stresses and Their Effects', The Welding Institute, Cambridge, 1981, pp 1-4.

36. Y. Akita, T. Yada, K. Sakai and N. Iino, "Brittle Fracture Initiation in a Residual Stress Field", Paper C2, Conference on Practical Application of Fracture Mechanics to Pressure Vessel Technology, London, 3-5 May, 1971, Inst. Mech. Eng.

37. C.L. Formby and J.R. Griffiths, "The Role of Residual Stress in the Fracture of Steel", International Conference on Residual Stresses in Welded Construction and Their Effects, The Welding Institute, London, 1977, pp 359-374.

38. G.R. Egan, "The Application of Fracture Toughness Data to the Assessment of Pressure Vessel Integrity", Paper II-74, Second International Conference on Pressure Vessel Technology, San Antonio, Texas, 1-4 October, 1973, ASME.

39. L.G. Taylor and R.A. Farrar, "Metallurgical Aspects of the Mechanical Properties of Submerged-Arc Weld Metal", Welding and Metal Fabrication, Vol 43, No. 4, May 1975, pp 305-310.

40. R.E. Dolby, "Review of Work on the Influence of Nb on the Microstructure and Toughness of Ferritic Weld Metal", IIW Doc. IX-1175-80, 1980.

41. J. Tenkula and V.K. Heikkinen, "The Effect of Parent Plate Aluminum on the Properties of Two-Wire Submerged-Arc Weld Metal in Nb-Microalloyed Steel", Welding Research Int'l, Vol 8, No 5, 1978, pp 369-377.

42. C.H. Entrekin, "Effect of Stress Relief on Microalloyed Weld Metal Impact Properties", Welding J., Vol 62, No. 8, August 1983, pp 197 s - 203 s.

43. J. Vrtel, "The Aging of Steel - Methods of Evaluating Strain-Aging Susceptibility", Technical Digest, 3, 1967, pp 163-169.

44. R.E. Dolby, "Factors Controlling Weld Toughness - The Present Position, Part 2, Weld Metals", The Welding Institute, Report No. 14/1976/M, 1976.

45. J.G. Garland and P.R. Kirkwood, "Towards Improved Submerged Arc Weld Metal", Metal Construction, Vol 7, No. 5 and 6, May/June 1975, pp 275-283, 320-330.

46. A.J. Rubin, J.H. Gross and R.D. Stout, "Effect of Heat Treatment and Fabrication on Heavy-Section Pressure-Vessel Steels", Welding J., Vol 38, No. 4, April 1959, pp 182 s - 187 s.

47. R.C. Cochrane, P. Terry and J.G. Garland, "Root-Run Toughness in Multi-Pass Manual Metal-Arc Weld", Welding and Metal Fabrication, Vol 44, No. 4, May 1976, pp 316-320 and Vol 44, No. 6, July/August 1976, pp 439-448.

48. A.A. Wells and F.M. Burdekin, "Effects of Thermal Stress Relief and Stress Relieving Conditions on the Fracture of Notched and Welded Wide Plates", British Welding J., Vol 10, No. 5, May 1963, pp 270-276.

49. W. Soete, "One-Run Versus Multi-Run Welds", Welding J., Vol 50, No. 3, March 1971, pp 127 s - 136 s.

50. T. Toyooka and K. Terai, "On the Effects of Postweld Heat Treatment", Welding J., Vol 52, No. 6, June 1973, pp 247 s - 254 s.

51. M.G. Dawes, "Fracture Initiation in Weld Metals", Welding and Metal Fabrication, Vol 40, No. 3, March 1972, pp 95-104.

52. B.J. Brindley and J.T. Barnby, "Dynamic Strain Aging in Mild Steel", Acta Metallurgica, Vol 14, December 1966, pp 1765-1780.

53. B. Houssin, G. Slama and P. Moulin, "Strain Aging Sensitivity of Pressure Vessel Steels and Welds of Nuclear Reactor Components", in 1st Int'l Seminar on "Assuring Structural Integrity of Steel Reactor Pressure Vessel", L.E. Steele and K.E. Stahlkopf (Eds.), Berlin (West), Germany, August 1979, pp 57-67.

54. W.C. Leslie and R.L. Rickett, "Influence of Aluminum and Silicon Deoxidation on the Strain Aging of Low-Carbon Steels", J. Metals, Trans. AIME, August 1953, pp 1021-1031.

55. R.E. Reed-Hill, "Physical Metallurgy Principles", 2nd Ed., D. Van Nostrand Company, New York, 1973.

56. J.M. Wheatley and R.G. Baker, "Mechanical Properties of a Mild Steel Weld Metal Deposited by the Metal-Arc Process", British Welding J., Vol 9, No. 6, June 1962, pp 378-387.

57. R.A. Farrar, L.G. Taylor and E.M. Harrison, "Effect of Stress Relieving on Fracture Properties of Submerged-Arc Welds of C-Mn Steels", Metals Technology, Vol 6, No. 10, October 1979, pp 380-389.

58. S.S. Tuliani, "The Role of Manganese in Submerged-Arc Weld Metal", Conference on 'Welding Research Related to Power Plant Engineering', CEGB, Marchwood, 1972, pp 232-246.

59. G. Oates, "Effect of Temperature and Strain Rate on Cleavage Fracture in a Mild Steel and a Low-Carbon Manganese Steel", J. Iron and Steel Inst., Vol 207, March 1969, pp 353-357.

60. D.A. Curry, "Cleavage Micromechanisms of Crack Extension in Steels", Metal Science, Vol 14, No. 8 & 9, September 1980, pp 319-326.

61. A.A. Griffith, Phil. Trans. R. Soc., Vol A221, 1920, p 163.

62. E. Smith, Proc. Conf., "Physical Basis of Yield and Fracture", Inst. Phys. and Phys. Soc., Oxford, 1966, p 36.

63. D.A. Curry and J.F. Knott, "The Relationship Between Fracture Toughness and Microstructure in the Cleavage Fracture of Mild Steel", Metal Science, Vol 10, No. 1, January 1976, pp 1-6.

64. D.A. Curry and J.F. Knott, "Effects of Microstructure on Cleavage Fracture Stress in Steel", Metal Science, Vol 12, No. 11, November 1978, pp 511-514.

65. R.O. Ritchie, W.L. Server, and R.A. Wullaert, "Critical Fracture Stress and Fracture Strain Models for the Prediction of Lower and Upper Shelf Toughness in Nuclear Pressure Vessel Steels", Met. Trans. A., Vol 10A, October 1979, 1557-1570.

66. T.C. Lindley, "The Effect of Pre-Strain and Strain-Aging on the Cleavage Fracture of a High-Manganese Steel", Acta Metallurgica, Vol 14, December 1966, pp 1835-1839.

67. J.D.G. Groom and J.F. Knott, "Cleavage Fracture in Prestrained Mild Steel", Metal Science, Vol 9, 1975, pp 390-400.

68. D.E. McRobie and J.F. Knott, "Effects of Strain and Strain Aging on Fracture Toughness of C-Mn Weld Metal", Materials Science and Technology, Vol 1, May 1985, pp 357-365.

69. O.E. Lepik, "Strain Aging Behaviour of Weld Metals Tested in the Darlington Leak Before Break Material Test Program", Ontario Hydro Research Division Report, to be issued.

TABLE 1

RESULTS OF COD TESTS ON 1 IN (25 mm) SQUARE SPECIMENS OF AS-DEPOSITED, STRAIN AGED AND THERMALLY STRESS-RELIEVED WELD METALS (REFERENCE 51)

Weld metal reference letter	Process	ASTM Classn.	Testing Temp. °C	Critical COD					
				As-deposited		Strain-aged		Strain-aged and stress-relieved	
				in. $\times 10^{-3}$	mm $\times 10^{-3}$	in. $\times 10^{-3}$	mm $\times 10^{-3}$	in. $\times 10^{-3}$	mm $\times 10^{-3}$
(e)	Manual metal-arc	E7018	−50	>28	>71	3	8	>28	>71
(f)		E7018G	−70	>28	>71	5	13	6	15
(k)		E8016C.1	−70	>28	>71	2	5	24	61
			−90	>25	>64	1	2·5	3	8
			−20	22	55	10	25	>25	>64
(s)		E6018	−40	28, 10	71, 25	7	18	24	61
			−60	10	25	4	10	20	51
			−80	14, 3	36, 8	<1	2	21	53
			−100	7	18	—	—	18, 11	46, 28
(m)	Electroslag		+40	>28	>71	3	8	—	—
			−18	25	64	<1	1	17	43
(o)	Flux-cored wire CO_2		−20	17	43	—	—	10	25
			+18	21	53	4	10	7	18

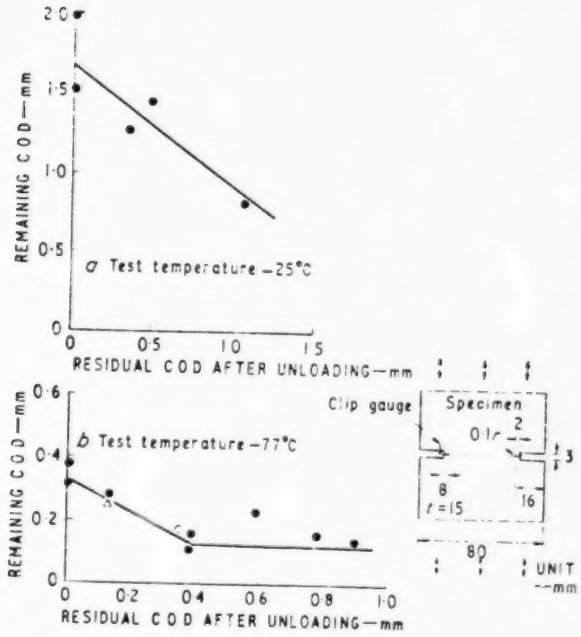

Figure 1: **Effect of Residual Stress or Residual COD on the Remaining COD at (a) -25°C; (b) -77°C. (Reference 36).**

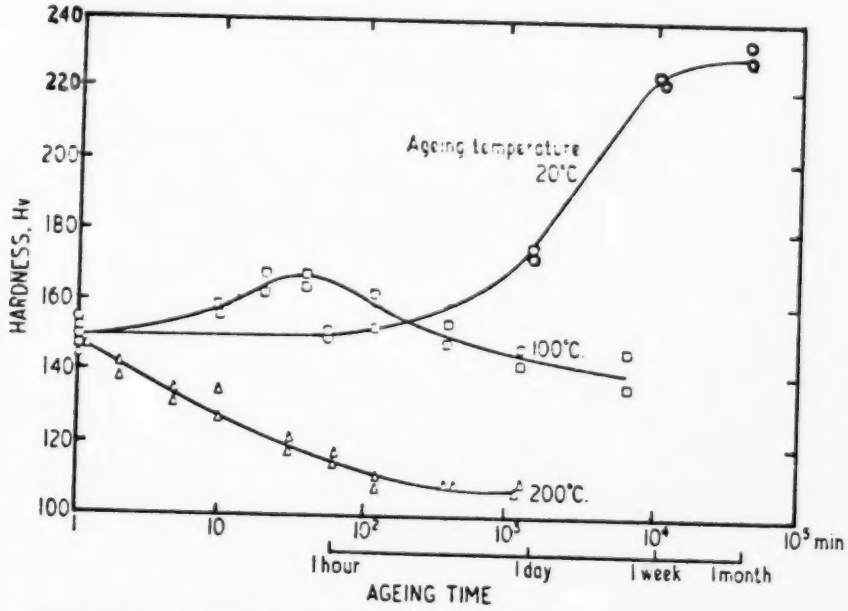

Figure 2: **The Hardness of Rimming Carbon Steel vs Aging Time at 20, 100 and 200°C. (Reference 43).**

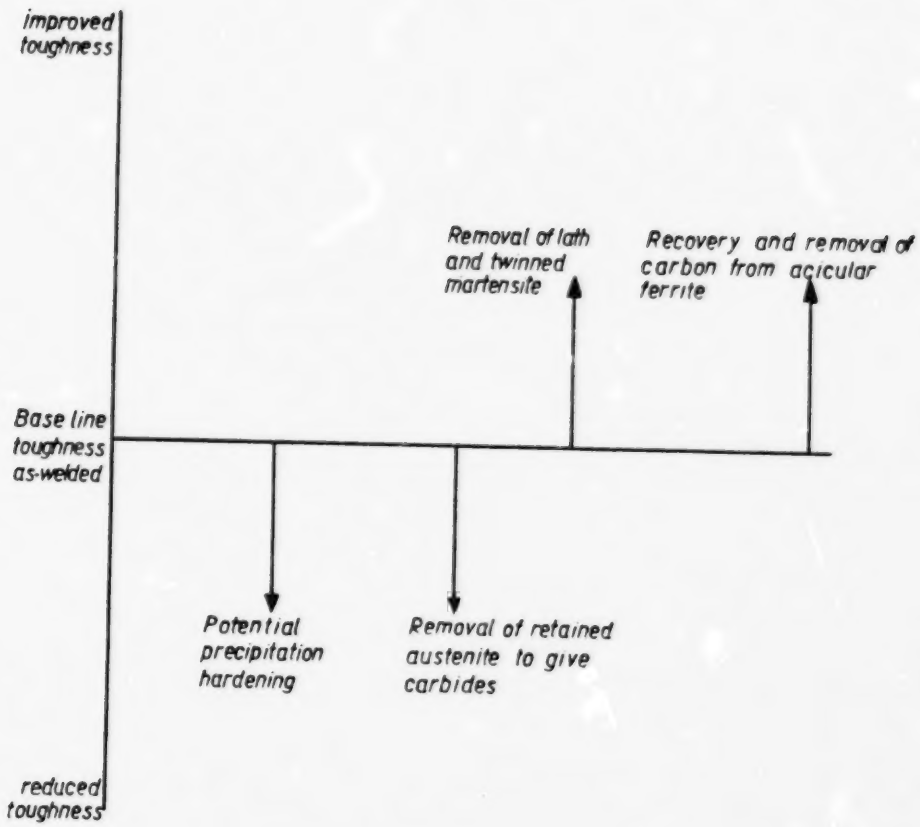
improved toughness
Removal of lath and twinned martensite
Recovery and removal of carbon from acicular ferrite
Base line toughness as-welded
Potential precipitation hardening
Removal of retained austenite to give carbides
reduced toughness

Figure 3: **Schematic Diagram Showing the Various Metallurgical Changes That Can Take Place During PWHT. (Reference 44).**

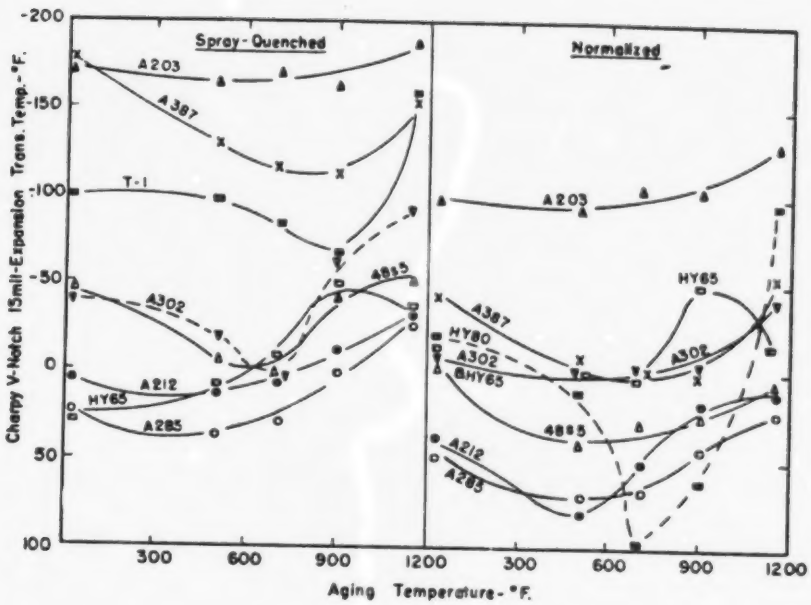

Figure 4: **Effect of Prestraining and Reheating at Various Stress Relieving Temperatures on Notch Toughness (A-285 and A-212 are C-Mn Steels). (Reference 46).**

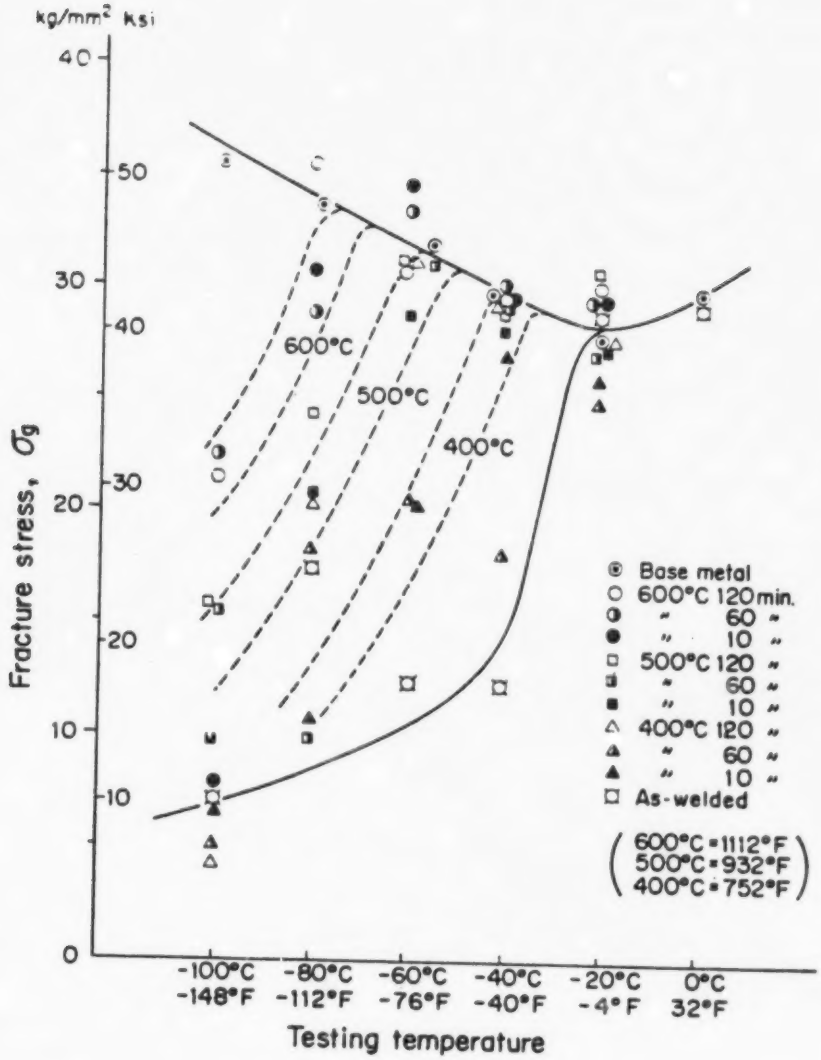

Figure 5: **Recovery of Fracture Stress After Postheating of a C-Mn Steel. (Reference 50).**

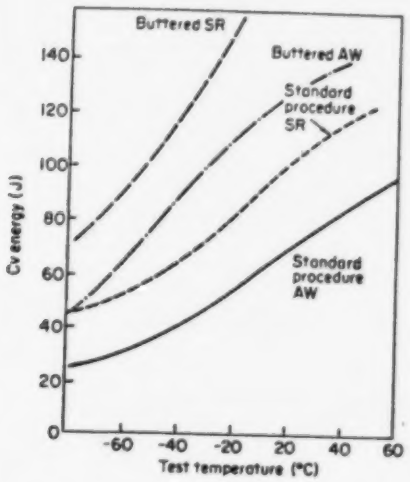

Figure 6: **The Influence of PWHT on the Charpy V-Notch Impact Strength of the Root-run in Multi-pass MMA welds in Q.T. C-Mn-Nb-V Steel. AW-As-Welded; SR-Stress Relieved. (Reference 47).**

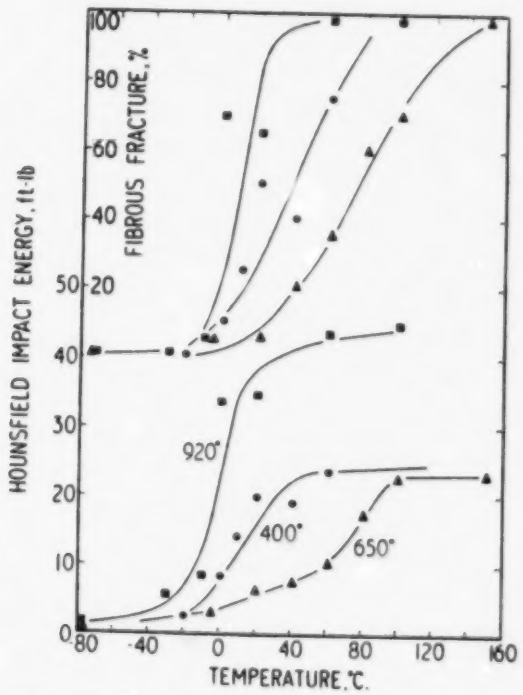

Figure 7: **Energy and Fibrosity of C-Mn Weld Metal (Electrodes of Class E6XX) Reheated for 2 h. (Reference 56).**

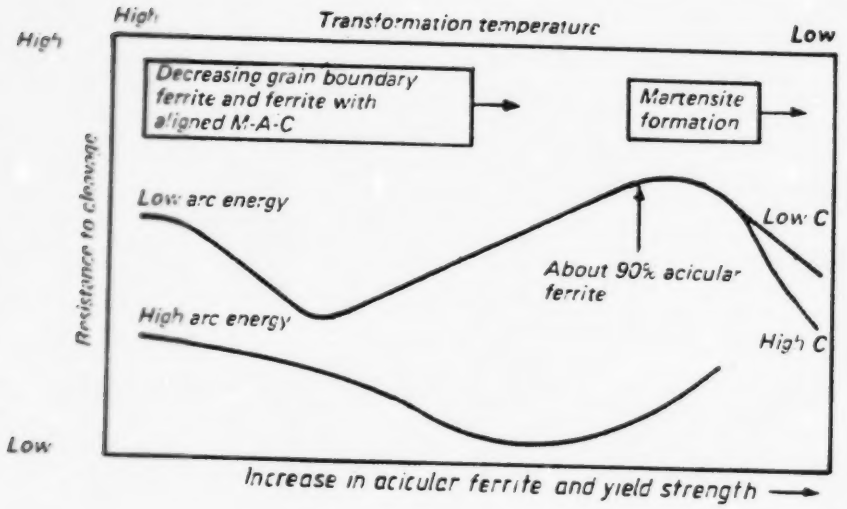

Figure 8: **Schematic Diagram Showing Relationship Between Microstructure and Cleavage Resistance of High and Low Arc Energy Welds (Reference 8).**

DISTRIBUTION

--Mr. M.R.	Allen	Design & Development - Gen., H14G27
--Dr. D.	Mills	Director of Research
Mr. M.L.	Aggarwal	Design & Development - Gen, H14F24
Miss B.L.	Kee	Design & Development - Gen, H9C4
Mr. M.J.	Kozluk	Design & Development - Gen, H14E23
Mr. M.L.	Lau	Design & Development - Gen, H14C22
Mr. B.W.	Manning	Design & Development - Gen, H14E25
--Mr. J.T.	Martin	Design & Development - Gen, H14A21
Mrs. N.D.	Martin	Design & Development - Gen, H14B21
--Dr. D.	Sidey	Technical & Training Services, A8H7
Mr. L.F.	Robertson	Technical & Training Services, A7H8
--Dr. J.A.	Baron	Metallurgical Research
Dr. M.A.	Clark	Metallurgical Research
Dr. R.G.	Fleck	Metallurgical Research
Dr. W.H.S.	Lawson	Metallurgical Research
Dr. B.	Mukherjee	Metallurgical Research
Mr. D.A.	Scarth	Metallurgical Research
Mr. M.J.	Tinkler	Metallurgical Research
Mr. M.L.	Vanderglas	Metallurgical Research
Mr. A.C.	Wallace	Metallurgical Research
Mr. O.E.	Lepik (2)	Metallurgical Research
Dr. E.T.C.	Ho (2)	Metallurgical Research

--Executive Summary Only

Editorial Section (2)
Public Reference Centre
Office Copies (3)
Extra Copies (3)